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The significance of local plasticity for the crack initiation process during very high cycle fatigue of high strength steels

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Abstract

The present paper deals with experimental results on the fatigue damage of grade SAE 318 LN austenitic-ferritic duplex steel and a modified SAE 4150 tempering steel in the very high cycle fatigue (VHCF) regime. Particular attention was paid to the relationship between the crystallographic orientation of individual grains and grain patches that exhibit slip band formation, fatigue crack initiation and growth. Therefore, electrolytically polished bulk specimens have been fatigued under fully reversed loading in an ultrasonic fatigue testing machine while the surface has been observed in-situ by an optical microscope. The specimens were carefully analyzed by means of scanning electron microscopy (SEM) in combination with automated electron back-scatter diffraction (EBSD). In case of the austenitic-ferritic duplex steel under VHCF loading conditions, slip band formation is limited to the softer austenite grains. Once being formed, the bands generate high stress concentrations where they impinge the austenite-ferrite ($\gamma-\alpha$) phase boundaries, eventually, leading to the crack initiation. In case of the tempering steel crack initiation is caused by slip band formation within the martensitic laths structure. It was found that in both materials microcracks are sensitive to changing crystallographic orientations when crossing a grain or phase boundary.

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Keywords: duplex stainless steel; tempering steel; fatigue crack; VHCF; EBSD

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1. Introduction

The assessment of the fatigue life of cyclic loaded structures is usually based on Wöhler data. Here the fatigue life is determined by the amplitude level which determines furthermore the effective damage mechanism. For example, a LCF loaded specimen shows an early formation of slip bands, which leads to initiation and formation of long fatigue cracks [1]. The decrease of the macroscopic strain amplitude increases the fatigue life, but the microscopic strain amplitude can reach a critical value due to the elastic anisotropy of the microstructure, which results in the formation and propagation of microstructurally short fatigue cracks. In the VHCF range, the initiation and propagation of fatigue cracks is mainly influenced by the microstructure. Therefore, a large scatter can be observed in the fatigue life data. Life-determining factors are size, shape and distribution of non-metallic inclusions and the ability of grain and phase boundaries to block cyclic slip or microcracks. Duplex stainless steels exhibit a good combination of high strength, ductility and corrosion resistance [2]. The martensitic steel SAE 4150 is used for high strength power transmission parts where a high ductility is needed. Both steels are used for applications which imply cyclic loading. To ensure the reliability of such machines the knowledge about the fatigue behavior is essential. The present paper gives some evidences about the fatigue damage mechanism for ferritic austenitic duplex stainless and martensitic steels in the VHCF regime.

2. Experimental Procedure

The following investigations are focused on the role of microstructural parameters on the fatigue behavior. Therefore, a precise characterization of the material is needed. Substantial changes on the microstructural parameters are controlled by varying heat treatment parameters, which are given in Table 1 for both materials SAE 4150 and SAE 318 LN.

Table 1: Chemical composition (wt. %) and heat treatment parameters for the SAE 318 NL and SAE 4150.

Material	C	Cr	Ni	Mo	Mn	N	P	S	Fe
SAE 318 NL	0.02	21.9	5.6	3.1	1.8	0.19	0.023	0.002	bal.
SAE 4150	0.48	1.00		0.18	0.71		0.013	0.010	bal.
SAE 318 NL	grain coarsening:1250°C (4h); cooling down to 1050°C (1K/min); water quench								
SAE 4150	quenching and tempering: 850°C (0.5h) quenched in oil; 540 - 680°C (1h) air								

The materials have been characterized by means of EBSD to determine the crystallographic orientations, grain and phase distribution. The prior austenite grain size of the SAE 4150 is about 12 µm, the length of the martensitic laths is varying between 5 to 20 µm. With respect to the duplex steel, the grain coarsening heat treatment resulted in a microstructure consisting of 50 % austenite and 50 % ferrite, with a mean grain size of 33 µm for austenite and a mean grain size of 41 µm for ferrite. The distribution of crystallographic orientations on a microscopic scale are determining the macroscopic properties, e.g., yield strength. Figure 1a and b show the crystallographic orientation of every individual grain in transversal direction in dependency to the rolling direction, which is perpendicular to the image plane. In case of the duplex steel in Fig. 1a a pronounced <100> orientation along the rolling direction is found. This investigation reveals a Kudjumov-Sachs crystallographic relationship between the {111} planes in austenite and the {110} planes in ferrite, which results from the phase transformation from ferrite to austenite during the heat treatment. The loading axis during the fatigue tests is parallel to the rolling direction. In Fig 1b the inverse pole figure (IPF) mapping of the SAE 4150 steel is given; the rolling direction is again perpendicular to the image plane. The measurement reveals the martensitic microstructure and prior austenite grains. These grain patches show a certain misorientation to each other. It is expected that these areas cause a local elastic anisotropy which induce compatibility stresses. The geometry of the specimens is given in Fig 1c, after lathing and milling the specimens were electro-polished in the gauge length to observe the surface of the specimens by means of light and electron microscopy in combination with EBSD.

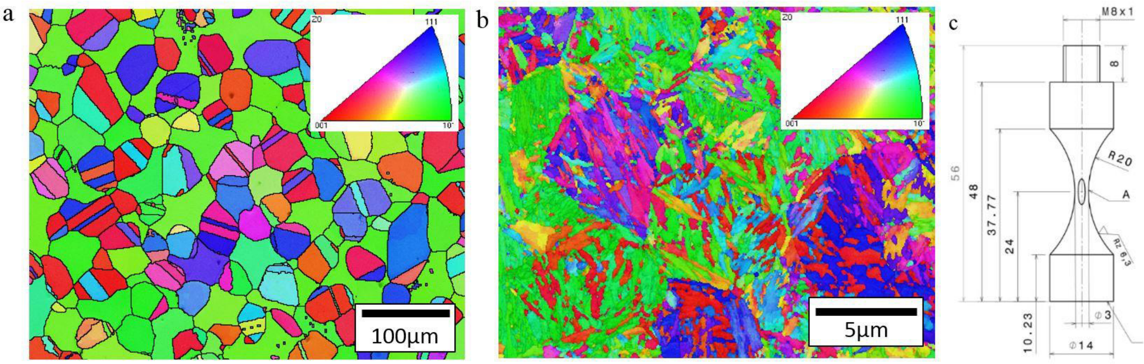

Figure 1. (a) Inverse Pole Figure (IPF) mapping of the SAE 318NL in transversal direction; (b) IPF mapping in TD of SAE 4150; (c) geometry of the fatigue specimen for the ultrasonic fatigue testing equipment

For cyclic loading an ultrasonic fatigue testing equipment designed and produced by BOKU Vienna was used, with an average frequency of f=20 kHz and an load ratio of R=-1. To prevent heating of the specimens during cycling, compressed air cooling is used. The observation of the fatigue damage within the machine is done with an optical far field microscope Questar QM100. The microscope is focused on the shallow notched area in the middle of the gauge length of the specimen shown in Fig 1c. The microscope is continuously taking pictures of the surface, thus a tracking of the crack growth is possible.

3. Results

3.1. Duplex stainless steel SAE 318 NL

As it was found in certain studies for the SAE 318 NL, cf. [3], a pronounced slip band formation in the austenite grains takes place during the tests. This behavior is shown for the austenite grains in Fig 2a and b. By taking continuously images of the fatigue samples during VHCF cyclic loading, it was found that the first slip bands can be observed during the first thousand cycles within the austenite phase. These grains show a continuously increasing number of activated slip bands during the tests. This means that local irreversible plastic slip of the SAE 318 NL has generally to be accepted for fatigue applications also in the VHCF regime on a microscopic scale. Furthermore, these slip bands impinge the austenite ferrite phase boundary. Two scenarios can be observed: (i) a pile up is formed at the phase boundary (ref. Fig 2a) or (ii) dislocations can be transferred over the austenite ferrite phase boundary into the ferrite grain by compatible slip systems (ref. Fig 2b). Generally, it seems that microcrack initiation for the duplex steel at VHCF loading amplitudes is depending on the amount of accumulated micro strain caused by the local plastic deformation of austenite grains at the phase boundaries (cf. Fig 2b). The cracked specimen shown in Fig. 3a was loaded with an stress amplitude of $\Delta\sigma/2$=400 MPa up to $1.6\cdot 10^6$ cycles until the test was stopped. By means of the observation of the specimen surface the area of crack initiation was identified to be at a $\gamma-\alpha$ phase boundary between the grains 1 and 3. Afterwards, the notched area of the specimen was investigated by EBSD to calculate the corresponding slip systems and Schmid factors for the grains in Fig 3a. The austenite grain 1 shows a continuously increasing number of slip bands during the test, whose Schmid factor was determined to be M_s=0.46 for the slip planes in (1;1;1;¯) direction. The slip traces on the specimen expand throughout the austenite grain 1, passing the phase boundary into the ferrite grain 2. This implies that in this case slip transmission across the phase boundary plays an important role for the crack initiation process. Therefore, the individual slip planes for the austenite grain 1 and ferrite grain 3 have been calculated and plotted into Fig 3a to prove compatibility conditions of slip planes. It is obvious that the orientation of the slip planes allow a slip transmission from the fcc (1;1;1;¯) plane across the phase boundary to the bcc (01;¯1) plane.

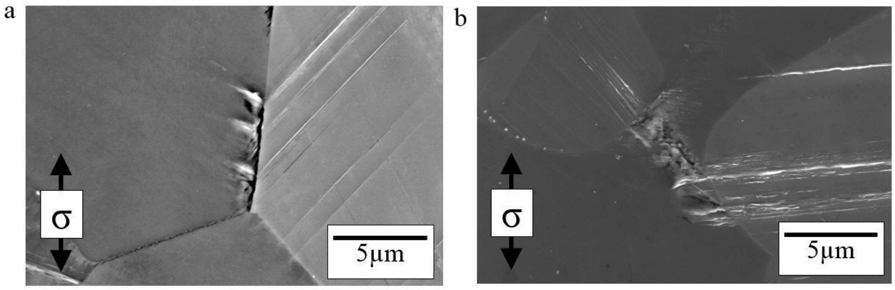

Figure 2. (a) Irreversible plastic deformation at austenite ferrite phase boundary; (b) local plastic deformation induced microcrack initiation

The optical surface observation allows the measurement of the surface crack length vs. the number of cycles for the left and right crack front in Fig 3a, which only corresponds to the first 50µm of the final crack length. To show the individual crack propagation rate, the crack length was measured separately for the left and right crack tip. It was found that the propagating crack follows within the first $1.3\cdot10^6$ cycles the phase boundary in the left direction until it reaches grain 5. In the other direction it took the same number of cycles to overcome the grain boundary between grain 1 and grain 2 to reach grain number 4.

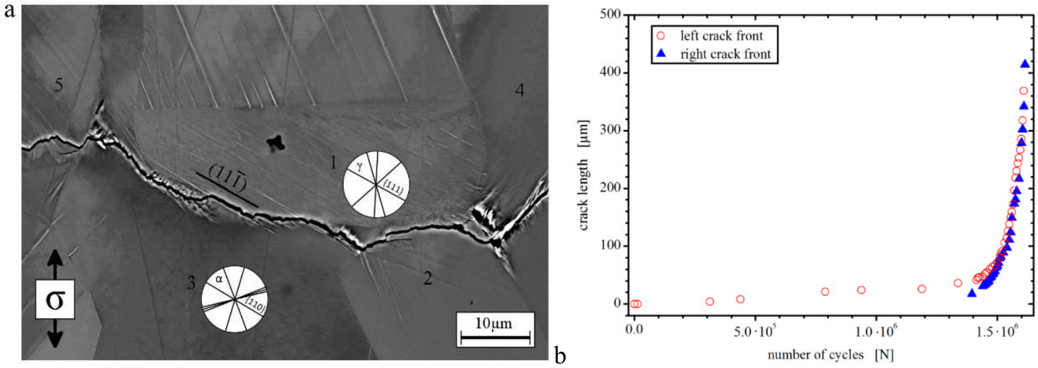

Figure 3. (a) Crack initiation site of a fatigue sample loaded at $\Delta\sigma/2$=400MPa, $1.6\cdot10^6$ cycles; (b) Crack length vs. the number of cycles for the left and right crack front.

The crack propagation starts to accelerate after crossing the phase boundaries to the grains 4 and 5, respectively. The data reveals that crack initiation takes place during the first 10^5 numbers of cycles even for VHCF loading amplitudes. The analysis of the tested set of samples shows, that for specimens which have almost reached 10^8 numbers of cycles microcrack initiation and propagation occurs. It was found that microcracks initiated at austenite ferrite phase boundaries and microcrack propagation starts in the first ferrite grain. Infinite fatigue life is related to effective obstacles that can slow down or stop microcrack propagation, for instance grain and phase boundaries.

3.2. Tempered steel SAE 4150

The investigation of the tempered steel SAE 4150 has been focused as well on the development of local plastic deformation and crack initiation on a microscopic scale caused during VHCF. The characterization of the microstructure of the fatigued samples by SEM and EBSD allows identifying the role of crystallographic orientations in relation to the mechanical behavior within the microstructure. It was found that the slip bands follow certain crystallographic orientations which seem to correlate with the direction of the martensitic laths (cf. Fig. 4). In case of the crack shown in Figure 4, the crack initiated by local plastic deformation caused by the elastic anisotropy of different oriented martensitic patches. Around position α1 a remarkable number of slip band exists, the orientation of the martensitic laths is along the favorable (101;¯) plane, which shows in this case the highest Schmid factor. The orientation of the martensitic laths for the areas α2 and α3 is also marked with dashed lines. The calculation of the possible slip planes reveals again a good agreement between the orientation of the laths along the {011} plane, however the Schmid factor for the patches α2 and α3 is lower than in the case of α1.

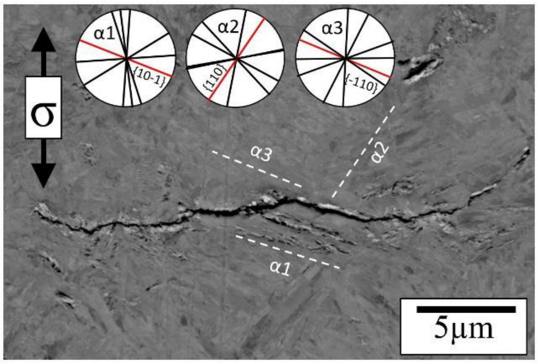

Figure 4. Crack initiation site of a SAE 4150 fatigue sample (Δσ/2=400MPa).

Hence, it can be concluded that the patches α2 and α3 deform elastically, while the orientation of the (101;¯) plane in α1 gives a high Schmid factor, which leads to plastic slip. This scenario is probably be assisted by the rise of additional stresses cause by the anisotropy of the involved martensitic patches, which finally lead to microcrack initiation along the point of contact of the patches α1, α2 and α3.

3.3. Finite-Element-Modelling

Based on the preliminary work of Huang [4] a model allowing the implementation of crystal plasticity was implemented into a Finite-Element-model that also accounts for elastic anisotropy. The model (realized by using ABAQUS) was applied to microstructures of the materials studied to predict crack initiation sites. The plastic deformation is determined by a velocity-dependent flow law, where plastic deformation on a slip system is dependent on the resolved shear stress [4]. The description of the cyclic stress-strain behavior is accounted for by a kinematic law for strain-hardening. An example of the application of the FE model on the duplex steel is given in Figure 5. The microcracks (see arrows in Figure 5a) were initiated at a stress amplitude of Δσ/2=365 MPa.

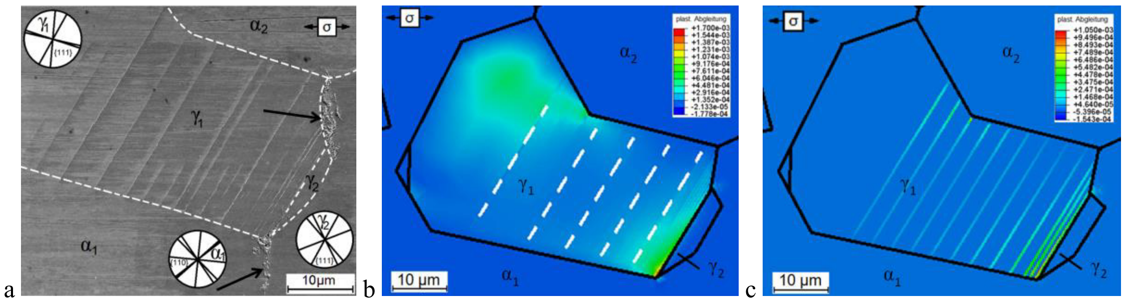

Figure 5: (a) intercrystalline und transcrystalline fatigue crack initiation (arrows) in the ferrite, introduced by local stress intensities caused by slip bands within the austenite; (b) plastic slip on the slip system with the highest resolved shear stress; (c) plastic slip on discrete slip bands.

The emanation process of the slip bands within the austenite grain γ_1 leads to stress intensities at the phase boundaries and in the neighboring ferrite grain α_1. Figure 5b shows the overall displacement on the slip system in the austenite grain γ_1, leading to pronounced slip band formation. To analyze the stress distribution at the phase boundaries, discrete slip bands have been implemented within the highly stressed austenite grain (cf. Figure 5c), according to the results of Figure 5b. These slip bands are activated when the local friction stress τ_{fr} is exceeded. Meanwhile, the residual areas of the austenite grain are deformed only elastically. This approach allows to calculate a global stress value, which is not sufficient to initiate any plastic deformation within the ferrite phase. Accordingly, the endurance limit for a certain area of the microstructure can be estimated. Details of the model are described elsewhere [5]. Figure 6a shows a detail of the EBSD map from Fig 1b of the tempered steel SAE 4150, which was implemented into the model introduced above. In Figure 6b the result for the strain calculation is given, showing the accumulated slip on all {110}<111> slip systems. The prior austenite grain boundary in Figure 6b is marked with a white dashed line. A significant amount of irreversible slip was found between the grains 12 and 3, just opposite of the prior austenite grain boundary the amount of plastic slip remains on a descent level.

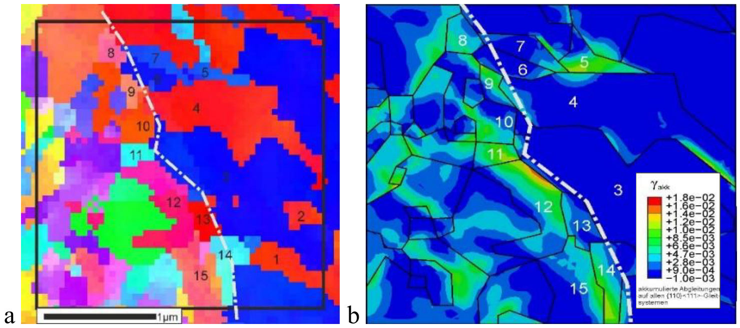

Figure 6. (a) Crystal orientation data obtained by EBSD; (b) accumulated slip on all {110}<111> slip systems.

The presented examples for the steels SAE 318 NL and SAE 4150 demonstrate that even at very low macroscopic plastic strain amplitudes in the VHCF regime the microstructural anisotropy generates significant stress intensities which determine the fatigue life.

4. Conclusions

It was shown that the fatigue life in the VHCF regime of high strength steels is related to the amount and distribution of local plastic deformation within the microstructure. For both materials tempered steel and duplex stainless steel, it was found, that the crystallographic orientation relationship on the slip planes plays the governing role. Under fatigue loading, the formation of slip bands was observed already on the first thousand cycles, leading to local stress concentrations at the phase and grain boundaries. In case of the SAE 4150, it was found that the elastic anisotropy of varying orientated martensitic patches is leading to intense irreversible slip along the prior austenite grain boundaries. These areas were identified experimentally by light microscopy and numerically by the finite element method as critical crack initiation sites.

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